MICROPLASTICITY AND ENERGY DISSIPATION IN VERY HIGH CYCLE FATIGUE

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Abstract This paper present the ongoing DISFAT project financially supported by the French National Agency ANR and its white program. It aims at a deeper understanding of mechanisms leading to crack initiation in metals and alloys in Very High Cycle Fatigue (VHCF). In contrast to conventional fatigue tests, detecting irreversible microstructural changes using cyclic stress-strain curves analysis is no more possible, at least with conventional measurement systems, since the macroscopic behavior is quasi-elastic. To cope with this problem, we propose to use the energy signature of the deformation mechanisms. The main challenge of this project results from the fact that the energy and mechanical manifestations as well as the microstructural “evidences”of the mechanisms of interest involve very low signal. In this paper, we explain our strategy and present some results on fcc and bcc single-phase polycrystals.

1 INTRODUCTION

The DISFAT project (“Microplasticity and energy dissipation in very high cycle fatigue”) is an ongoing French project financially supported by the French National Agency ANR and its white program. It aims at a deeper understanding of mechanisms leading to crack initiation in metals and alloys in Very High Cycle Fatigue (VHCF). The VHCF regime is associated with stress magnitudes lower than the conventional fatigue limit and as a result, numbers of cycles higher than 10^9. The present project is a fundamental research project. However, it is motivated by industrial problems. Some mechanical components, such as pistons, rotating axes, in the transportation and more generally engineering industry, have been designed previously using fatigue resistance data at lower numbers of cycles ($<10^7$ cycles ; the regime of High Cycle Fatigue, HCF) whereas they must endure oscillating loads for a number of cycles higher than $>10^9$ cycles (Bathias and Paris, 2004). In addition, many experiments prove that failure occurs at stress amplitudes lower than the conventional fatigue limit. There is also growing demand for the development of robust life prediction and damage assessment methodologies that allow the safe extension of service lifetimes for transportation and power generation systems beyond their original design lifetimes; the methodologies must also allow for the design of new efficient systems intended to operate at even longer lifetimes. These requirements motivate the need to understand the fundamental mechanisms of fatigue in the VHCF regime, correspondingly, to explore novel methods for the characterization of fatigue behavior at these very long lifetimes. VHCF has been a subject of growing interest in recent years giving rise to four International Conferences on Very High Cycle Fatigue (the first one was organized by Pr C. Bathias in Paris in 1998, next Vienna in 2001, Kyoto in 2004, Michigan in 2007, Berlin in 2011).

In the High Cycle Fatigue (HCF), much work, including analyzing and modeling proposed by Pr. H. Mughrabi (e.g. Wang and Mughrabi, 1984; Sommer et al, 1998) has been performed, and a fair understanding has emerged over the years. Cyclic strain localization in Persistent Slip Bands (PSBs) is generally found and accepted as the critical onset of fatigue damage in the HCF regime in ductile single-phase materials and also in many precipitation-hardened alloys. Once formed, PSBs, accompanied by the development of extrusions and notch-like deepenings, commonly referred to as intrusions, lead to cracks. Basically, cyclic strain localization in PSBs represents fatigue mechanisms, characterized by well-defined threshold amplitudes. Fatigue damage by cyclic slip can only occur when elementary forward slip displacements during one half-cycle are not perfectly reversed during a subsequent half-cycle. The cyclic slip is thus commonly referred as partially irreversible. Here, the term “irreversible” is not used in the sense of thermodynamics. It simply refers to cyclic slip processes that cause permanent or irreversible microstructural changes in the bulk leading to damaging notch-peak geometries at the surface. The irreversible microstructural changes are mainly related to dislocation processes. PSB and dislocation structure in the matrix depend on the grain orientation, grain size and more generally on the local mechanical state. In addition, the ability of screw dislocations to cross slip and annihilate is known to be a key-parameter for the dislocation structuration processes. Do the fatigue mechanisms that are become effective in the VHCF range are simply those, mentioned above, that are known to be active in the HCF range? Most current studies have focused on the VHCF behaviour of high-strength materials containing microstructural heterogeneities such as inclusions and referred as type II materials (Mughrabi, 2002). An important finding was that, in the transition from HCF to VHCF, the origins of fatigue failure changed from surface to interior “fish-eye” fracture at non-metallic inclusions (e.g. Bathias and Paris, 2004; Mughrabi, 2006). What about ductile single-phase materials which do not contain hard nonmetallic inclusions? For ductile single-phase materials, referred as type I materials, only few investigations mainly on copper (Stanzl-Tschegg et al, 2007; Weidner et al, 2010; Stanzl-Tschegg and Schönbauer, 2010; Krupp et al, 2011; Höppel et al, 2011) are available regarding the shape of the fatigue life curve and the damage evolution. From these experiments, it is suggested that crack initiation will take place at the surface owing to the accumulation of very small but irreversible plastic deformation over very large number of load changes (Mughrabi, 2002). However, there is no possibility for the formation of a pronounced PSB structure at very low stress amplitudes below the PSB threshold. Therefore, it is necessary to conduct systematic investigations of type I materials after cyclic deformation in the VHCF regime in order to clarify the role of the above mentioned mechanisms and parameters.

In this project, we will focus on both face-centered (fcc) and body centered (bcc) cubic single-phase metals, displaying planar or wavy slip. These materials are good candidates because their HCF behavior has been largely studied in literature and we will be able to refer to this knowledge. A special attention will be paid to the role of grain size and orientation and of the initial surface state following the work of Mareau (2007) and Favier et al (2010) in HCF regime. In addition to the question of the material dependency of fatigue mechanisms addressed above, studying the VHCF range requires developing specific experimental equipments and methodologies. To perform experiments up to a very high number of cycles in a reasonable time, ultrasonic equipment at a testing frequency of 20 kHz will be used. It takes only fourteen hours to perform 10^9 cycles at testing frequency of 20 kHz whereas it takes one hundred days at 100 Hz. In contrast to conventional fatigue tests, detecting irreversible microstructural changes using cyclic stress-strain curves analysis is no more possible, at least with conventional measurement systems, since the macroscopic behavior is quasi-elastic. To cope with this problem, we propose to use the energy signature of the deformation mechanisms. The temperature increase of the specimen will be measured during the cyclic loading. The dissipated energy will be deduced from temperature field measurements by solving the heat equation accounting for exchanges losses and material conductivity. This calorimetric method has been successfully developed for conventional fatigue at testing frequencies ranging from 10-50 Hz (Boulanger et al, 2004; Meneghetti, 2007; Chrysochoos et al, 2008; Doudard et al, 2010; Connesson et al, 2011). In this project, it should be adapted to 20 kHz cycling loading. From an industrial point of view, ultrasonic fatigue testing and dissipation estimates during cycling loading may appear as rapid methods to select or to class materials with regard to their fatigue properties. However, it addresses the issue of the frequency impact on the fatigue response. To clarify these aspects and the reliability of these methods as well as for better understanding of physical mechanisms, multi-scale observations of the material of interest will be performed to follow the microstructural changes at the surface and/or in the bulk.

The project brings together four academic partners with complementary fields of expertise and equipements (PIMM – Arts et Métiers ParisTech, LEME - Université Paris X, LMGC - Université Montpellier II, LSPM – UPR CNRS). A technical center (CETIM Senlis) is also associated to transfer the results of this fundamental work to industrial applications. Most of participating researchers have worked in the HCF fatigue field and Professors Bathias and Mughrabi who contributes to the project provided pioneering and well-known works in the VHCF range. In this paper, we first introduce background on VFCF. Then, we explain our experimental procedure. Finally, some results are presented in order to illustrate our methodology.

2 BACKGOUND

Definition of the VHCF regime

During the last 30 years, studies of metal fatigue to extended fatigue lives of more than 10^7 cycles to failure (Awatani et al, 1975; Willertz, 1980; Hessler et al, 1981; Bathias, 1999), were mainly focused on the shape of the fatigue live diagram (also referred as S-N curves, S: stress, N: number of cycles to failure or Wöhler curves) with special emphasis to the so-called “fatigue limit”. Figure 1 represents the typical fatigue life diagram characterized by three or four stages according to the authors. These four stages can be characterized as follows (from Mughrabi, 2002):

I. Finite fatigue life regime extending from the low cycle fatigue (LCF) to the high cycle fatigue (HCF) range. The fatal cracks initiate most commonly at the surface.

II. Conventional HCF fatigue limit. It is characterised by some form of cyclic strain localisation threshold. It corresponds to the PSB threshold for ductile single metals.
III. Regime of finite very high cycle fatigue (VHCF) life. Mughrabi suggests distinguishing two classes of metallic materials: type I materials are pure ductile single-phase metals containing no extrinsic internal defects; type II materials contain internal defects such as non-metallic inclusions/dispersoids. For type II materials, cracks originate from internal defects and lead to the so-called fish-eye fracture. On the opposite, for type I materials, crack initiation will take place at the surface owing to accumulation of very small but irreversible microstructural changes over very large number of load. This regime is below the PSB threshold.
IV. VHCF fatigue limit. The presence of a fatigue limit related to stage IV is still debated. Some authors (see e.g. Bathias and Paris, 2004) suggest a continuously decreasing stress-life response and prefer introducing a fatigue strength associated with a given number of cycles rather than a fatigue limit. Others (see e.g. Mughrabi, 2002) suggest the presence of a fatigue limit below the irreversibility threshold characterized by non-damaging slip or negligible slip irreversibility and infinite fatigue life.

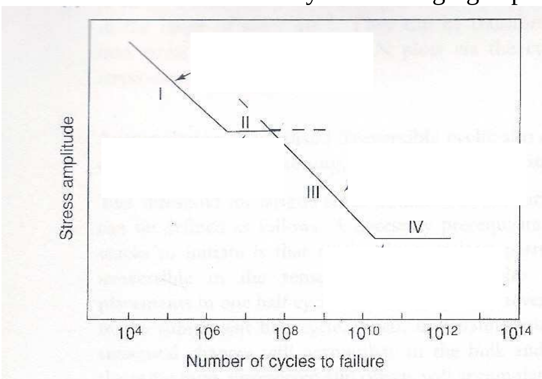


Figure 1. Schematic illustration of proposed shapes of multi-stage**e** S-N curve. (a) after Nishijiama and Kanazawa (1999)

Irreversible microstructural changes prior to crack initiation

This question of the existence of a fatigue limit is clearly linked to mechanisms prior to crack initiation. In particular, irreversible microstructural change is obviously required to generate rupture of the material in ductile metals but is probably not sufficient. In the HCF regime, the role of persistent slip bands (PSBs) on the initiation of fatigue cracks in metals has long been recognized. In fcc metals, PSBs in which the plastic strain is localized have a well-defined dislocation structure, the so-called ladder (wall) dislocation structure, characterized by a length separating walls of dislocation dipoles. The ability of cross slip and annihilation impacts the dislocation accumulation processes (Laird, 1996; Lukas, 1996; Desprès, 2004). For example, referring to the investigations of Thompson and Backofen (1971) on wavy-slip materials such as Cu and Al but also on planar-slip α-brass, it was shown that for wavy slip materials and grain sizes between some ten and several hundred micrometers, homogeneous tangle of dislocations are found and no considerable grain size influence on fatigue properties in the HCF region is visible while for planar slip materials, planar slip bands form and the grain size influences significantly the fatigue life curve. Mughrabi et al (1979) showed that a necessary prerequisite for cyclic strain localization in PSBs is that the so-called PSB threshold amplitude be exceeded. However, in the VHCF regime, well below the PSB threshold, marked cyclic slip localization occurred in intense slip bands after very large numbers of cycles (Stanzl-Tschegg et al, 2007; Weidner et al, 2008, Weidner et al, 2010 ; Stanzl-Tschegg and Schönbauer, 2010; Krupp et al, 2011; Höppel et al, 2011). Does a limit of this cyclic slip localization, more generally labeled as microstructural irreversibility “evidence”, exist? What are the conditions to activate these microstructural irreversibility “evidences”? Do similar mechanisms involved in HCF occur in VHCF but with a lower kinetics? These are the main questions addressed in this project.

Methods for the characterization of VHCF fatigue behaviour

The need to understand the fundamental mechanisms of fatigue in the VHCF regime requires exploring novel methods for the characterization of fatigue behavior at these very long lifetimes. As mentioned above, ultrasonic fatigue testing at frequency of 20 kHz brings advantages to provide fatigue limit and near threshold crack propagation data within a reasonable time. However, the question of the frequency effect on the fatigue response is addressed. Firstly, owing to the high frequency, heating of specimens is observed during ultrasonic fatigue tests. This heating in some materials can reach 200°C (Wagner et al, 2006; Ranc et al, 2008) and consequently modifies the material properties. To prevent heating of specimens, compressed air was used as a cooling medium (Bathias, 2006). Intermittent testing can also be a solution (Stanzl-Tchegg et al, 2007). Secondly, the frequency effect on the fatigue response may be due to intrinsic parameters (strain rate sensitivity) and extrinsic parameters (corrosion). Actually, the trend is to say that in fcc materials, where the cyclic flow stress is weakly dependent on strain rate, little effect is expected (Morrissey and Nicholas, 2006), while in bcc materials sensitive to high strain rate, the frequency effect could be more important (Tsutumi et al, 2006; Setowaki et al, 2011). However, recent papers (Engler-Pinto et al, 2007) have shown a frequency effect on fcc Al-Si and Al-Si-Cu alloys between 75 Hz and 20 kHz which depends on the strength of the alloys.

The temperature variation in a specimen during a cyclic loading may also be used to detect internal changes within the material. Indeed, it depends on the heat sources developed within the material (sources associated with dissipation of the mechanical energy into heat and sources associated with the coupling phenomena (e.g. thermoelasticity)). Obviously, the passage from temperature to heat sources requires the estimate of heat losses by diffusion in the specimen and of heat transfer with the surroundings by convection and radiation. In order to obtain information on the energy signature of the deformation mechanisms, it is necessary to derive, by an inverse method, the associated heat sources from the temperature field measurement. Many authors used temperature measurements during fatigue tests to understand, for example, the effect of the stress amplitude on the temperature rise (e.g. Luong, 1995; Fargione et al, 2002; Curà, 2005). More recently, some research groups developed energy approaches based on the dissipation assessments (Boulanger et al, 2004; Meneghetti, 2007; Chrysochoos et al, 2008; Doudard et al, 2010; Connesson et al, 2011). Thermal and calorimetric studies were all carried out in the HCF range but not in the VHCF regime.

3 EXPERIMENTAL PROCEDURE

3.1 Materials and specimens

To understand the mechanisms of interest, our strategy is to analyse the response of metals and alloys having “simple” microstructure and deformation mechanisms but also controlled initial states, with special attention paid to the surfaces (with reduced roughness and residual stress state, labeled as “ideal” state surface). From the existing knowledge of cyclic deformation mechanisms, we choose to study the role of two intrinsic properties of solid crystals: the tendency to wavy or planar slip and the lattice friction resistance. In this project, we propose to study two classes of ductile single-phase metals with fcc and bcc structures. In both classes, we used a quasi pure metal and alloys in order to change the ability to cross slip. Concerning fcc metals, pure copper and Cu-Zn (α-brass) alloys are good candidates. While pure Cu is known to deform with wavy slip, α-brass Cu-Zn displays planar slip. Here, we only present results concerning a commercial polycrystalline copper CuOF 99.99% supplied by Griset Company. Concerning bcc metals, low Fe-C were used and here, we present results of Armco iron. The mean grain size in both materials was approximately 50 µm.

In order to ease surface observation, a new hourglass shaped specimen with flat faces was designed (Fig.2). The specimen dimensions were determined so that all the parts, such as transmission and amplification pieces, vibrate at a resonance frequency of 20 kHz in tension-compression (Phung et al, 2011; Wang et al, 2011). After mechanical and electrolytic polishing, the specimen surfaces were mirror finish without any residual stresses. The stress distribution along the specimen axis (calculated numerically) is presented in Figure 2. The stress is concentrated in the middle of the gauge part of the specimen and reduced toward the specimen ends. This geometry allows to systematically study the material response at all desired stress amplitudes with one single specimen.

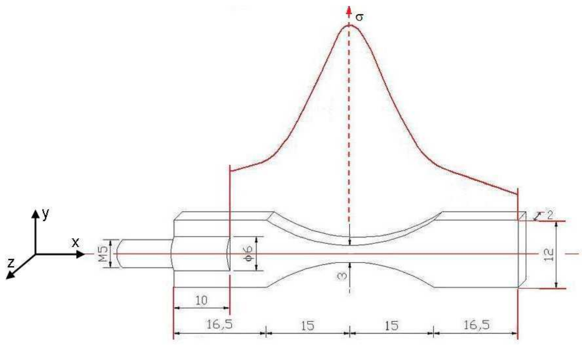


Figure 2. Ultrasonic fatigue plate specimen and distribution of stress along the specimen axis for CuOF 99.99%.

3.2 Fatigue tests

To reach VHCF regime within a reasonable time, ultrasonic fatigue technique at a testing frequency of about 20 kHz has been used at very small stress amplitude (lower than the VHCF fatigue limit). Assuming an elastic behavior, the strain ratio, (minimum strain over maximum strain) was R = -1. One side of specimen was painted in black so that the heating change during loading will be captured by an infrared camera. From these temperature measurements, the intrinsic dissipation was determined using a 1D thermal model (see below). The fatigue experiments were periodically interrupted after prescribed numbers of loading so that the development of plastic slip markings on the other side surface was observed by optical and electronic microscopy.

3.3 Model of heat diffusion

A numerical model was built to estimate the distribution time course of intrinsic dissipation from temperature measurement fields during the fatigue test. We focused on the longitudinal distribution of heat sources within the specimen gauge part. A 1D calorimetric analysis has been justified assuming in a first approximation a uniaxial tension-compression stress state. From the heat equation:

$$\rho C \dot{T} - k\Delta T = s \tag{1}$$

where T is the temperature, ρ the mass density; C the heat capacity; k the thermal conduction coefficient. s(x,y,z,t) symbolizes the volume of heat source. Following Boulanger's hypotheses (Boulanger et al, 2004), the 1D diffusion equation for a non-constant cross-section can be written as (Doudard et al, 2010):

$$\frac{\partial\theta(x,t)}{\partial t} + \frac{\theta(x,t)}{\tau^{1D}(x)} - \frac{k}{\rho C}\left(\frac{\partial^2\theta(x,t)}{\partial x^2} + \frac{\partial\theta(x,t)}{\partial x}\frac{S'(x)}{S(x)}\right) = \frac{s(x,t)}{\rho C} \tag{2}$$

with $\theta = T - T°$ is the temperature change, $T°$ is the room temperature and τ^{1D} is a time constant term which characterizes the heat losses through lateral surfaces of the specimen:

$$\tau^{1D}(x) = \frac{\rho C \cdot S(x)}{2h(e + l(x))} \tag{3}$$

where e is the specimen thickness, $l(x)$ is its width with respect to x. We note $S(x) = e \cdot l(x)$ the cross-section at this point. The mean dissipation over several thousand cycles was solely estimated (the thermo elastic sources being not considered here regarding the test frequencies and the adiabatic character of the thermoelastic processes over a complete cycle duration (50 µs)). In the above equations, the unknown parameter is the heat transfer coefficient h. The method used to identify h is presented in the following section.

3.4 Heat loss time constant identification

The heat transfer coefficient is determined for each test, from thermal field measurements when the fatigue loading is stopped and the temperature of specimen returns to thermal equilibrium. More precisely, the initial temperature was considered as the temperature when the load was stopped. During the thermal return to equilibrium, no heat source occurs. Thermal measurements $\theta^{\exp}$ applied to each end of the specimen by Dirichlet method enabled us to know the boundary conditions. The unknown heat transfer coefficient h was well chosen to satisfy these conditions:

$$\begin{cases} \dfrac{\partial\theta(x,t)}{\partial t} + \dfrac{\theta(x,t)}{\tau^{1D}(x)} - \dfrac{k}{\rho C}\left(\dfrac{\partial^2\theta(x,t)}{\partial x^2} + \dfrac{\partial\theta(x,t)}{\partial x}\dfrac{S'(x)}{S(x)}\right) = 0 \\ \theta(x,t=0) = \theta^{\exp}(x,t=0) \\ \theta(\dfrac{-L}{2},t) = \theta^{\exp}(\dfrac{-L}{2},t) \end{cases} \tag{4}$$

As a result, h was found in ranges of 30 – 100 W/m^2/K. This result shows that the heat losses are cause by natural convection and also by an air flow above the specimen which aims at cooling the piezoelectric system.

4 SOME RESULTS

In this section, we present some results illustrating our works. Figure 3 shows the PSBs observed at 10^8 cycles on the specimen surface for CuOF 99.99 material and Armco iron. The stress amplitude was chosen a little bit higher than the PSB stress threshold for both materials estimated at 10^8 cycles. It is worth noticing that the PSB stress threshold depends on the number of cycles (Stanzl-Tschegg and Schönbauer, 2010; Phung et al, 2011). This is the reason why we arbitrary chose 10^8 cycles. For both materials, we observed very few PSBs on the surface specimen located in isolated grains. The features of slip markings were founde different. For CuOF 99.99%, PSB are straight, lonely, mainly located near the grain boundaries. For Armco iron, they are spread in the whole grain and seem to be wavier. The formation of PSBs induced a self heating of the specimen due to intrinsic dissipation. Figures 4a and 4b display the average temperature over the gauge length and several thousands of cycles and the corresponding intrinsic dissipation for various maximum stress amplitude versus the number of cycles for CuOF 99.99%. The temperature rises over cycles and never reaches a constant value. In other words, the temperature does not stabilize, showing an evolution of the heat balance and consequently of the microstructure. Despite a slight raise of the temperature at $\Delta\sigma/2$ = 45.9 MPa, the intrinsic dissipation increased very slowly with the number of cycles. It attained to 0.498 °C/s at 10^6 cycles and reach to 0.505 °C/s at 10^8 cycles. It means that the heat sources were higher than the heat losses and remained active along the cycles. However, no slip bands were observed on specimen surface up to 10^8 cycles at this stress range. At $\Delta\sigma/2$ = 51.5 MPa, the intrinsic dissipation increased slowly up to 10^7 cycles. No slip bands were either observed between 10^6 and 10^7 cycles. At 10^8 cycles, a clear increase of the intrinsic dissipation was recorded. In this case, slip markings were observed on the specimen surface, the intrinsic dissipation was 1.322 °C/s. At higher stress amplitude, $\Delta\sigma/2$ = 56.8 MPa, the intrinsic dissipation increases with the number of cycles more rapidly than in previous cases. At $\Delta\sigma/2$ = 72.1 MPa, the intrinsic dissipation rises very fast with the number of cycles and reach to 7.477 °C/s at 10^6 cycles.

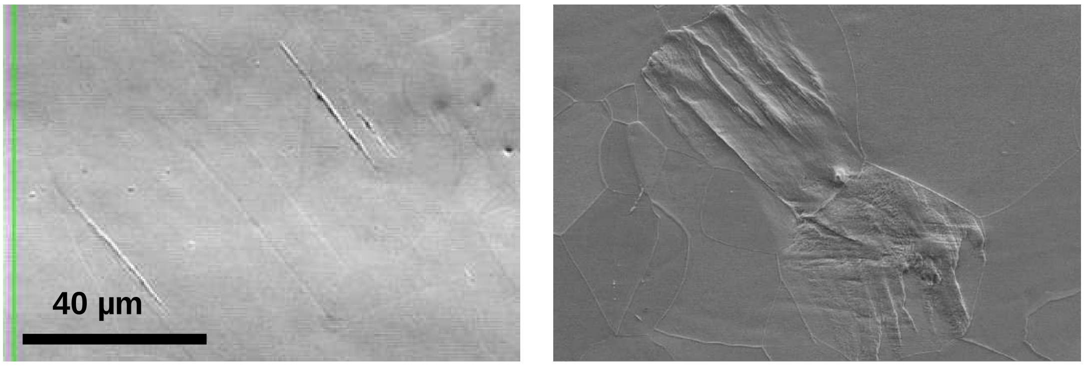
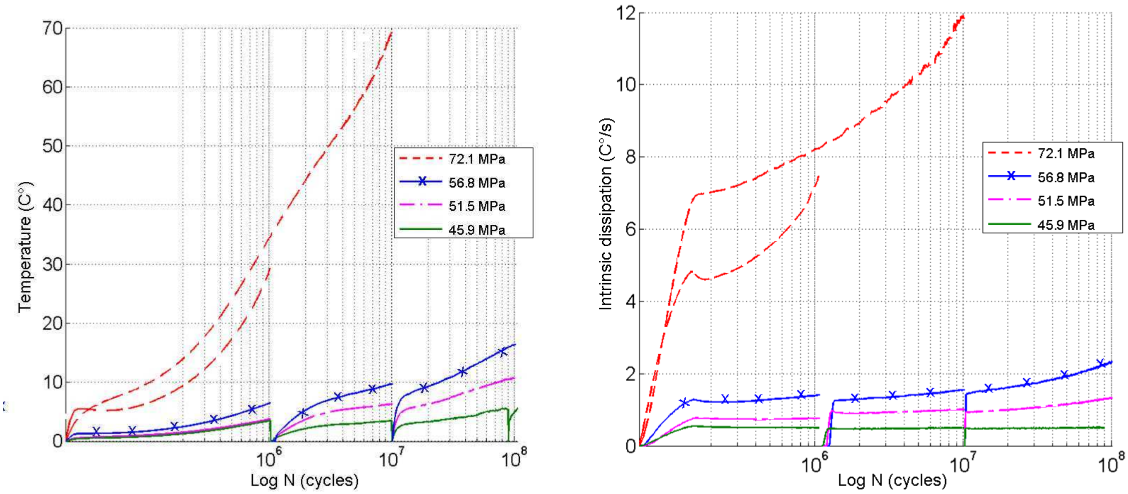


Figure 3: Persistent slip bands observed just below the PSB stress threshold at 10^8 cycles. (a) CuOF 99.99% at 68 MPa- (b) Armco iron at 95 MPa.



(a) (b)

Figure 4. (a) Average temperature and (b) Average intrinsic dissipation during fatigue test at different constant stress amplitude fatigue test for CuOF 99.99%.

In addition to average values, the intrinsic dissipation distribution can be estimated along the specimen axis (Fig. 5a). The intrinsic dissipation is concentrated in the middle of the specimen and decreases toward the ends. Hence, it is, as expected, related to the distribution of stress (Fig.2). Optical micrographs revealed that this zone display the highest quantity of slip markings (Fig. 5b). This result confirms that dissipation is related to microplasticity.

5 CONCLUSION

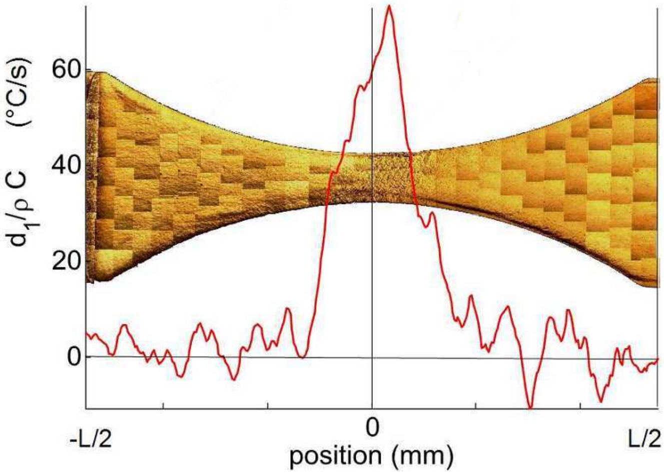
The DISFAT project aims at a deeper understanding of mechanisms leading to crack initiation in metals and alloys in Very High Cycle Fatigue (VHCF). The main challenge of this project results from the fact that the energy and mechanical manifestations as well as the microstructural “evidences”of the mechanisms of interest involve very low signal. As a result the objectives are twofold:

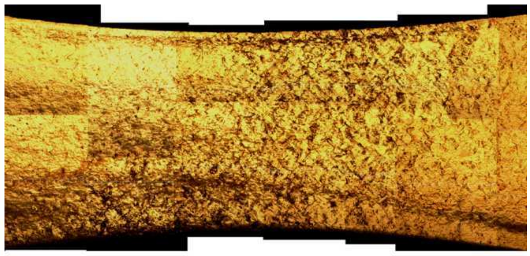
- exploring and developing novel techniques to characterize fatigue mechanisms in the VHCF range

- establishing (1) the impact of intrinsic properties of crystals such as the tendency to wavy slip or planar slip, the lattice friction and (2) the role of the surface on the cyclic plastic strain localization and crack initiation in the VHCF range.

The strategy consists in characterizing internal change of polycrystalline materials by coupling two methods: (1) observing the specimen surface change during cycling and (2) estimating the intrinsic dissipation from temperature measurements. A new hourglass specimen with flat faces has been designed to ease field analysis. First results show that:

- The PSB stress threshold depend on the number of cycles
- Slip markings associated with PSBs appear at the specimen surface in few and isolated grains for both materials. However, they are straight and located at the grain boundaries for CuOF 99.99% whereas they are wavy and spread over the grain for Armco iron.
- Field intrinsic dissipation along the specimen axis was estimated from thermal 2D measurements achieved on the flat hourglass specimen.
- Intrinsic dissipation is related to microplasticity activity characterized by slip markings: higher dissipation corresponds to higher quantity of slip markings. However, CuOF 99.99% and Armco iron dissipate energy even at stress amplitudes for which no slip markings were observed.





(a) (b)

Figure 5. Intrinsic dissipation along the specimen at $\Delta\sigma/2 =$ 72.1 MPa, 10^7 cycles (a) and a zoom at the middle of gauge specimen (b)

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